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NASA TECHNICAL MEMORANDUM

NASA TM X-52357

NASA TM X-52357

N67-39653

(ACCESSION NUMBER) (THRU)

26 (PAGES) 1 (CODE)

TMX-52357 (NASA CR OR TMX OR AD NUMBER) 17 (CATEGORY)

FACILITY FORM 602

APPLICATION OF A METHOD OF ESTIMATING HIGH TEMPERATURE LOW CYCLE FATIGUE BEHAVIOR OF MATERIALS

by G. R. Halford and S. S. Manson
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TECHNICAL PAPER proposed for presentation
at National Metal Congress sponsored by the
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NATIONAL AERONAUTICS AND SPACE ADMINISTRATION · WASHINGTON, D.C. · 1967

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E-4191

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ABSTRACT

Further study is made of a recently proposed method whereby tensile and stress-rupture properties may be used to estimate low-cycle, strain fatigue behavior in the creep range. The method is based primarily on the equation of universal slopes previously developed for predicting room-temperature fatigue behavior, and on a straightforward linear damage analysis of combined fatigue and stress-rupture. Fatigue life estimates are presented and compared with a large quantity of fatigue data for nickel-base alloys, high- and low-alloy steels, stainless steels, and aluminum-base alloys tested under laboratory conditions over a wide range of temperatures and cycling frequencies.

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INTRODUCTION

Low-cycle, strain-controlled fatigue resistance is now recognized as an important materials characteristic in some high-temperature systems such as gas turbines and nuclear reactors. The amount of elevated-temperature fatigue data generated with laboratory samples, although increasing, is still somewhat limited and data are seldom available for any given alloy under prescribed temperature and cyclic rate conditions. Since elevated temperature testing may also be rather time consuming and expensive, estimates of the performance of specific materials of engineering interest are very useful, particularly in the early stages of design and materials selection.

In a recent paper (1), we proposed a method for estimating isothermal, strain-controlled, low-cycle fatigue behavior of materials in the "creep range" from a knowledge of only elevated-temperature tensile and stress-rupture properties. The method was applied to forty sets of strain-controlled fatigue data on laboratory size samples obtained from a variety of sources over a wide range of temperatures and cyclic frequencies. Since that time, considerable additional data have been analyzed and these too are compared with estimates obtained by the proposed method.

We have chosen to devote our studies to total strain range-fatigue life behavior instead of to plastic strain range or elastic strain (stress) range alone. The reasons for this are first, that low-cycle fatigue tests are generally conducted under control of total strain range, and second, that field measurements of strain are in terms of total strain. Hence, design information is advantageously presented in these terms.

THE METHOD OF ESTIMATION

Only the salient features of the procedure for estimating fatigue life will be presented here, and the reader is referred to our earlier paper (1) for a detailed account of the assumptions and approximations involved in the derivation. Briefly, the method is used to estimate "lower bound", "average", and "upper bound" fatigue lives for a given total strain range. The upper bound and average lives, respectively, are always taken as constant factors of ten and two times the lower bound life.

In most cases, the lower bound life is given by 10 percent N_f and can be determined directly from the universal slopes equation proposed by one of the authors (2)

$$\Delta\epsilon_t = \frac{3.5\,\sigma_u}{E} N_f^{-0.12} + D^{0.6} N_f^{-0.60} \tag{1}$$

where

$\Delta\epsilon_t$ = Total strain range

N_f = Cycles to failure

σ_u = Ultimate tensile strength

E = Modulus of elasticity

D = True tensile ductility [determined from the percent reduction of area, R.A., by $\ln\left(100/(100 - \mathrm{R.A.})\right)$]

The tensile properties are determined at the elevated temperature of interest. Due to a dearth of specific data on the effect of strain rate on tensile properties at high temperatures, we have used properties determined at conventional strain rates. In the DISCUSSION section, we point out the potential merits of using tensile properties generated at strain rates comparable to the cyclic strain rates encountered during a strain-controlled fatigue test.

Extreme cases occasionally arise when the cyclic frequency and stress-rupture resistance of the material are low enough that the cyclic strain failures are especially time-dependent, and it becomes necessary to make an alternate estimation of the lower bound life according to

$$N_f' = \frac{N_f}{1 + \frac{k}{AF}(N_f)^{(m+0.12)/m}} \tag{2}$$

In this equation,

N_f' = Lower bound fatigue life

k = Effective fraction of each cycle for which the material may be considered to be subjected to the maximum stress. Ideally, this fraction depends upon several factors, but for reasons given in

Ref. (1), k is taken to be 0.3.

F = Frequency of stress application or the inverse of the cyclic period if dwell times are experienced, expressed in cycles per unit time.

A = Coefficient characterizing a time intercept of the stress-rupture curve of the material at test temperature. The curve of stress, σ_r, against rupture time, t_r, is assumed to be linear on logarithmic coordinates, and represented by the equation $\sigma_r = 1.75\ \sigma_u (t_r/A)^m$, so that A is the time intercept at an extrapolated value of $\sigma_r = 1.75\ \sigma_u$.

m = Slope of the linear log-log stress-rupture line (negative value).

N_f = Fatigue life calculated from equation (1).

Equation (2) represents a simplified attempt to quantitatively account for the combined fatigue and stress-rupture damage associated with the application of stress at high temperatures.

Figure 1 provides a simple criterion for judging whether the use of equation (2) is necessary. The computation need be made only if the point representing the coordinates (m) and (AF) that apply to the test conditions of the material lies above the curve. We have limited the application of equation (2) to cases where the stress-rupture slope (m) is steeper than -0.12. For slopes shallower than -0.12, accurate stress-rupture information is required before reliable fatigue life estimates can be made since equation (2) becomes sensitive to small variations in (m) and (A) in this region.

To summarize, estimates of elevated-temperature, low-cycle fatigue behavior are made as follows:

(1) Determine the <u>lower bound</u> of life, using either 10 percent N_f as computed from eq. (1) or N_f' as computed from eq. (2), whichever is lower.

(2) For <u>average</u> life, use twice the lower bound life.

(3) For the <u>upper bound</u> of life, use ten times the lower bound life. The application of these estimates to numerous sets of fatigue data is discussed later.

EXPERIMENTAL DATA

To further evaluate the method of estimation, additional high-temperature, strain-controlled, low-cycle fatigue data have been located in the literature (3-15); with the majority of the data coming from the recent Conference on Thermal and High-Strain Fatigue sponsored by the Institute of Metals (London). Over 75 sets of data are included, comprising nearly 600 fatigue tests in the life range from 10^1 to 10^5 cycles to failure. Test temperatures range from 932° F to 1650° F for the higher melting temperature alloys and from 300° F to 900° F for the aluminum alloys. Table 1 lists the materials, test conditions, and the pertinent tensile properties at the fatigue test temperatures. The tensile properties were presumed to have been obtained from conventional tests, since in some cases the authors did not cite complete information. In a number of cases, tensile data were not reported along with the published fatigue data, and hence, handbook and other reference source values were used as noted in the table.

The materials include nickel-base alloys, high- and low-alloy steels, stainless steels, and aluminum-base alloys. Ultimate tensile strengths ranged from about 2 to 160 ksi, and reductions of area from 4.5 to 99 percent. With the exception of a few tests in argon or vacuum, all tests were conducted in air environments at homologous temperatures (ratio of absolute test temperature to absolute melting point temperature) ranging from 0.43 to 0.82. Test frequencies were as low as 0.0077 cpm and as high as 300 cpm; with some tests involving dwell times at maximum strain up to 24 hours. All of the fatigue tests were conducted under strain control, either in plane bending or axial push-pull.

RESULTS OF APPLYING METHOD

The low-cycle fatigue data are plotted in figures 2 to 19 as total strain range versus cycles to failure on logarithmic coordinates. Unless noted in the figures, the tests were conducted in plane bending. Three curves are shown with each set of data, representing estimates made by the proposed method. The lowest curve is the estimate of the lower bound, using the rule associated with 10 percent N_f or N_f'; the middle curve is the estimate of average behavior; and the upper curve is the estimate of the upper bound. Continuous curves represent life estimates based on equation (1), whereas dashed curves denote that equation (2) was applied in making the estimates. Values of the stress-rupture slope (m) and the time intercept (A) used in the latter estimates are shown in the respective figures. In every instance, the criterion provided by figure 1 was checked to determine whether or not it would be necessary to base the estimates on equation (2). Since only eight sets of data required the use of equation (2), we have been unable to assess its validity to any great extent. In fact the dashed curves shown in figure 19(B) provide overly conservative estimates of the low-cycle fatigue behavior. As will be discussed later, closer estimates of fatigue life can be obtained in this case, and others, if tensile data generated at lower and more appropriate strain rates are used.

Examination of figures 2 to 19 reveals that qualitative estimates of the experimental results can be achieved by the proposed method. When applied to the present data, the method provided a lower bound for about 85 percent and an upper bound for about 95 percent of the data points. A measure of the degree of dispersion of the data from the estimated average behavior is given in Table 2. Tabulated is the percent of data included within various factors in life on either side of the estimated average lives. The agreement between the estimates and the data is reasonable, when it is recognized that: (i) the experimental results represent a wide variety of materials with divergent properties tested over a broad range of conditions; (ii) several different types of testing equipment and techniques were used; and (iii) only a limited number of easily determined tensile and stress-rupture properties were used in making the estimates.

Including the data previously analyzed (1), well over 100 sets of high-temperature, low-cycle fatigue data have been examined, and the agreement is sufficiently good that we have developed considerable confidence in applying the method to other materials and test conditions.

DISCUSSION

The method for estimating high-temperature, low-cycle fatigue behavior evolved from an initial consideration (16) of the potentially large cyclic life reductions brought about by the occurence of creepinduced intercrystalline cracking. Intercrystalline cracking was viewed as a means of reducing or bypassing the normal crack initiation stage of fatigue. However, other important factors such as oxidation, thermally-induced transformations, strain rate effects, etc. may well be involved, leading to a rather complex set of circumstances. Since the method does not consider these factors explicitly, more attention must be paid to them if more refined life estimates are desired.

The best procedures for refining the method are not yet clear; however, a few possible approaches were outlined in reference (1) as a guide for future study. One of these dealt with the effects of strain rate on the tensile properties and the stresses developed during cyclic deformation. At homologous temperatures above approximately 0.5, tensile strength and ductility can become highly strain rate sensitive (17), so that if low frequency (hence, low average cyclic strain rate) fatigue behavior is estimated from tensile data determined at significantly higher strain rates, good agreement might not be expected. As an example of the improved agreement that can be achieved by using low strain rate tensile data to estimate low frequency fatigue behavior, we have re-examined the 0.0077 cpm fatigue results of Forrest and Armstrong (15) on Nimonic 90 at 1600^{o} F shown in Figure 19(B). Forrest and Armstrong kindly supplied us with tensile properties generated at low strain rates approximately equal to the average cyclic strain rates ($\approx 2F\Delta\epsilon_t$) encountered in these low frequency fatigue tests. The estimated average life fatigue curve, CB, using these tensile data is shown in Figure 20. This estimate is considerably improved over the original (curve AB), which is shown again for comparison. All of the other low frequency (0.10 cpm) fatigue data of Forrest and Armstrong analyzed in this paper and in Ref. (1) have been re-examined in the above manner and the estimates of life have been improved or equalled in every case but one. It was not possible to re-evaluate any of the remaining fatigue data included in this paper, since tensile data at appropriate strain rates were not available.

By suggesting that strain rate effects be taken into account it is assumed that the cycle-dependent, pure-fatigue resistance is affected by strain rate to the same degree as the tensile properties. For example, if the tensile properties σ_u and D are reduced by decreasing the tensile strain rate, then the pure-fatigue resistance at an equivalent cyclic strain rate is expected to be reduced in accordance with that predicted by equation (1).

The above concept of strain rate dependent fatigue resistance is consistent with a parameter approach adopted by Coffin (18, 19, 20) for predicting the plastic strain fatigue behavior of rate sensitive steels. For temperatures near and below half the melting temperature, Coffin has demonstrated satisfactory agreement between predicted and observed fatigue data on low-carbon, strain-ageing steels. For temperatures above half the melting point, Forrest and Armstrong (15) have also incorporated strain rate effects into predictions of fatigue behavior. They used the equation, $\Delta\epsilon_p(N_f)^{0.5} = D/2$, where $\Delta\epsilon_p$ is the plastic strain range, to predict fatigue behavior for the Nimonic series of nickel-base alloys. The agreement for Nimonic 75 and Nimonic 90 was satisfactory for fatigue lives below about 1000 cycles. For Nimonic 105, fatigue lives were overestimated by factors of ten or more for cyclic lives above 100. As Forrest and Armstrong point out, the discrepancy between predicted and observed lives may be associated with a change in fatigue fracture mode that is not reflected in the tensile ductility, despite the fact that the strain rates in the tensile and fatigue tests were comparable.

We would like to suggest that at temperatures well into the "creep range" (homologous temperatures above approximately 0.5), stress-rupture damage may reduce cyclic lives beyond any reductions reflected by the effect of strain rate on the tensile properties. In suggesting that the influence of strain rate on tensile properties be considered when estimating fatigue behavior, it is nevertheless recognized that creep effects on fatigue life are important and should also be accounted for.

CONCLUDING REMARKS

The method we have proposed affords a simple way of estimating elevated-temperature, low-cycle fatigue behavior of materials from tensile and stress-rupture properties. Fatigue life estimates based upon the method have been compared with over 75 sets of high-temperature fatigue data on a variety of materials tested over a wide range of test conditons. The method provided lower bound fatigue lives for about 85 percent of the data, upper bound lives for approximately 95 percent of the data, and nearly 80 percent of the data fell within a factor of three on either side of the estimated average lives. The favorable agreement between the estimates and observed behavior suggests that it may be applied with confidence to other materials and test conditions. Such fatigue life estimates can be very useful, particularly in the early stages of design and materials selection. It is not intended, however, that these estimates eliminate the need for further experimental evaluation of the low-cycle fatigue behavior of materials at elevated temperatures. Rather, they are best suited for quick, approximate answers, and not for final design purposes.

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TABLE 1. - ALLOYS, TEST CONDITIONS, AND PERTINENT PROPERTIES USED IN ESTIMATES.

Alloy designation (as reported)	Test temper-ature, °F	Approximate homologous temperature	Test frequency, cpm	R.A., %	Tensile strength, σ_u, ksi	Elastic modulus, E, 10^3 ksi	Ref.
1197 Aluminum	300	0.46	10	95	4.0	[a]9.0	(3)
	600	.64	10	99	2.12	6.5	
1132 Aluminum	300	.46	10	80	8.5	[a]9.0	
	600	.64	10	99	5.41	6.5	
M-257 Aluminum (SAP)	600	.64	10	27	[a]17	[a]8.2	
	900	.82	10	23.0	9.1	[a]6.1	
Esshete 1250 (B1)	1110	.50	.5, 9.0	54	54.5	[a]22	(4)
Esshete 1250 (B2)	1110	.50	.5	55	51.5	[a]22	
Esshete 1250 (B3)	1110	.50	.5, 9.0	56	69.6	[a]22	
Esshete 1250 (B4)	1110	.50	2.0	63	60.0	[a]22	
A 286-Aged	1110	.50	.5, 2.0, 9.0	15	115	[a]23.1	
$2\frac{1}{4}$Cr-1 Mo Steel	1110	.48	1.0	80	[a]43	23.5	(5)
3 Cr-1/2 Mo Steel	932	.43	.5	81.7	78.0	[a]24	(6)
$2\frac{1}{4}$Cr-1 Mo Steel	1020	.45	.5, 9.0	60	65	[a]24	(7)
1 Mo Steel	950	.43	3.0	[a]77	[a]50	[a]24	(8)
1 Cr-1 Mo Steel	950	.43	3.0	[a]72	[a]55	[a]24	
12 Cr-1/2 Mo Steel	950	.47	3.0	[a]75	[a]75	[a]26	
1 Cr-1 Mo-1/4 V Steel (A1)	1050	.46	.1, 0.5, 1.0, 5.0, 10.0, 13.5	70	69.5	[a]24	(7)
1 Cr-1 Mo-1/4 V Steel (A2)	1020	.45	1.0, 10.0	60	72.5	[a]24	
1 Cr-1 Mo-1/4 V Steel (A5)	1020	.45	.5, 10.0	50	71.7	[a]24	
1 Cr-1 Mo-1/4 V Steel (A8)	1020	.45	1.0, 10.0	72	67.5	[a]24	
1 Cr-1 Mo-1/4 V Steel (A9)	1020	.45	.5, 10.0	70	56	[a]24	
1 Cr-1 Mo-1/4 V Steel (A11)	1020	.45	.1, 1.0, 10.0	68	54.5	[a]24	
1/2 Cr-1/2 Mo-1/4 V Steel (B1)	1020	.45	.5	70	60	[a]24	
1/2 Cr-1/2 Mo-1/4 V Steel (B2)	1020	.45	5.0	72	60.3	[a]24	
1/2 Cr-1/2 Mo-1/4 V Steel (B3)	1020	.45	1.0, 10.0	70	60	[a]24	
1 Cr-1 Mo-1/4 V Steel (A10)	1020	.45	1.0, 9.0	73	49.5	[a]24	
C-1/2 Mo Steel	1020	.45	5.0	72.5	47	[a]24	(9)
3/4 Cr-3/4 Mo-1/4 V Steel	1020	.45	5.0	78	57	[a]24	
1 Cr-1 Mo-1/4 V Steel (A)	1020	.45	.5, 9.0	77	60.7	[a]24	(10)
1 Cr-1 Mo-1/4 V Steel (B)	1020	.45	.5, 9.0	75	57.4	[a]24	
18/8 Stainless Steel	932	.46	.5	65.7	70.3	[a]24	(6)

[a]Handbook or other reference source data

TABLE 1. - Concluded. ALLOYS, TEST CONDITIONS, AND

PERTINENT PROPERTIES USED IN ESTIMATES.

Alloy designation (as reported)	Test temperature, ^{o}F	Approximate homologous temperature	Test frequency, cpm	R.A., %	Tensile strength, σ_u, ksi	Elastic modulus, E, 10^3 ksi	Ref.
304 Stainless Steel	1300 (Argon)	0.58	0.5	[a]52	[a]36	[a]20.5	(11)
	1500 (Argon)	.65	.5	[a]42	[a]20	[a]19	
	1600 (Argon)	.68	.5	[a]40	[a]16	[a]18	
20Cr-25Ni-0.7Nb Stainless (solution treat)	1200	.55	.5	40	70	22.5	(12)
20Cr-25Ni-0.7Nb Stainless (solution treat and aged)	1200	.55	.5	50	65	22.5	
20Cr-25Ni-0.7Nb Stainless (annealed)	1200	.55	.5	54.1	61.4	22.5	
20Cr-25Ni-0.7Nb Stainless (annealed)	1380	.61	.5	74.8	37.1	21.6	
316 Stainless Steel (A1 & A2)	1110	.52	.5, 9.0	58	51.3	[a]22	(4)
316 Stainless Steel (A3 & A4)	1110	.52	.1, 1.0, 2.0, 10	58	51.3	[a]22	
316 Stainless Steel (A4)	1200	.55	1.0	[a]53	[a]44	[a]21	
316 Stainless Steel	1500	.65	300	[a]45	[a]29	[a]19	(13)
316 Stainless Steel (A5)	1110	.52	.5, 1.0, 9.0	74	56	[a]22	(4)
316 Stainless Steel (A6)	1110	.52	2.0	74	60	[a]22	
316 Stainless Steel (A7)	1110	.52	2.0	73	54.6	[a]22	
316 Stainless Steel (A8)	1110	.52	1.0, 2.0	78	58.7	[a]22	
316 Stainless Steel (A9)	1110	.52	.5, 9.0	70	51.5	[a]22	
Udimet 700 ("E")	1400	.64	.67-1.8	31	155	23.6	(14)
Nimonic 75	1200	.57	.1	29.5	81.5	[a]25.7	(15)
	1380	.63	.1	42	53.1	[a]24.5	
	1600	.71	.1	62.5	28	[a]22.8	
Nimonic 105 (standard heat)	1380	.63	.1	18	151	[a]25.2	
	1600	.71	.1	32.5	99.5	[a]23.2	
Nimonic 105 (brittle heat)	1380	.63	10	15	141.5	[a]25.2	
	1600	.71	10	4.5	112.5	[a]23.2	
Nimonic 105 (ductile heat)	1380	.63	10	36.5	154	[a]25.2	
	1600	.71	10	43	103.5	[a]23.2	
Nimonic 90	1500	.67	10	13	96	[a]24	
	1600	.71	.0077	14	76	[a]23.2	
	1650	.73	10	22	61	[a]22.5	

[a]Handbook or other reference source data

TABLE 2. - DISPERSION OF EXPERIMENTAL FATIGUE LIVES FROM ESTIMATED AVERAGE LIVES

Factor in life above and below estimated average life	Percent of data included
1.5	36
2	59
3	80
4	89
5	94
6	96
8	97
10	98

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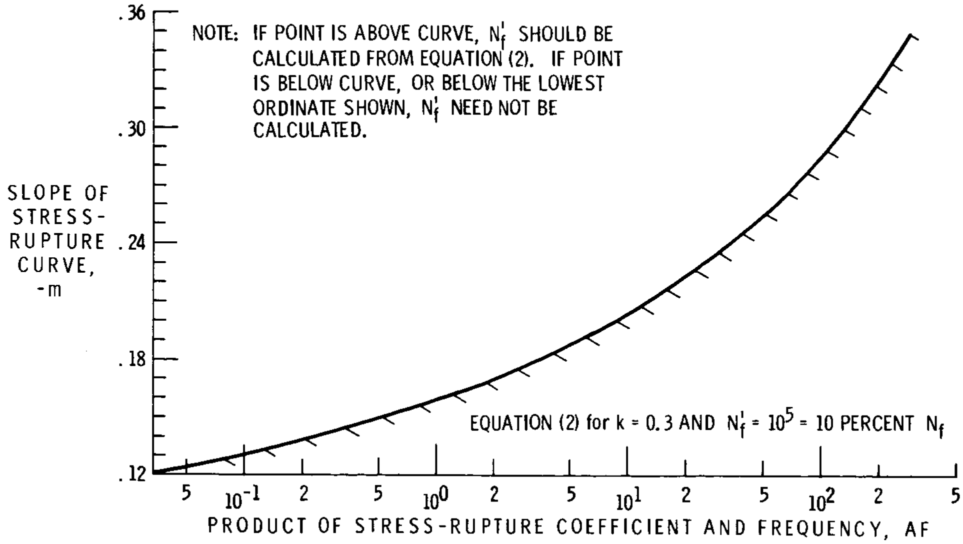

FIGURE 1. - CRITERION FOR ESTABLISHING WHETHER EQUATION (2) SHOULD BE USED IN ESTIMATING HIGH TEMPERATURE LOW CYCLE FATIGUE BEHAVIOR.

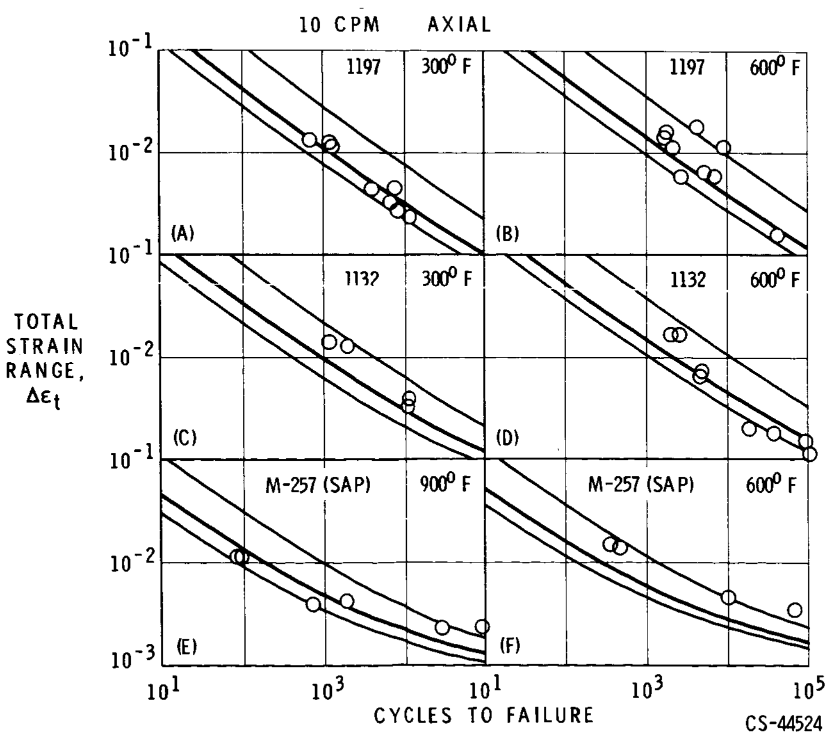

FIGURE 2. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [ALUMINUM ALLOYS] (A) - (F) ANDERSON AND WAHL (REF. 3).

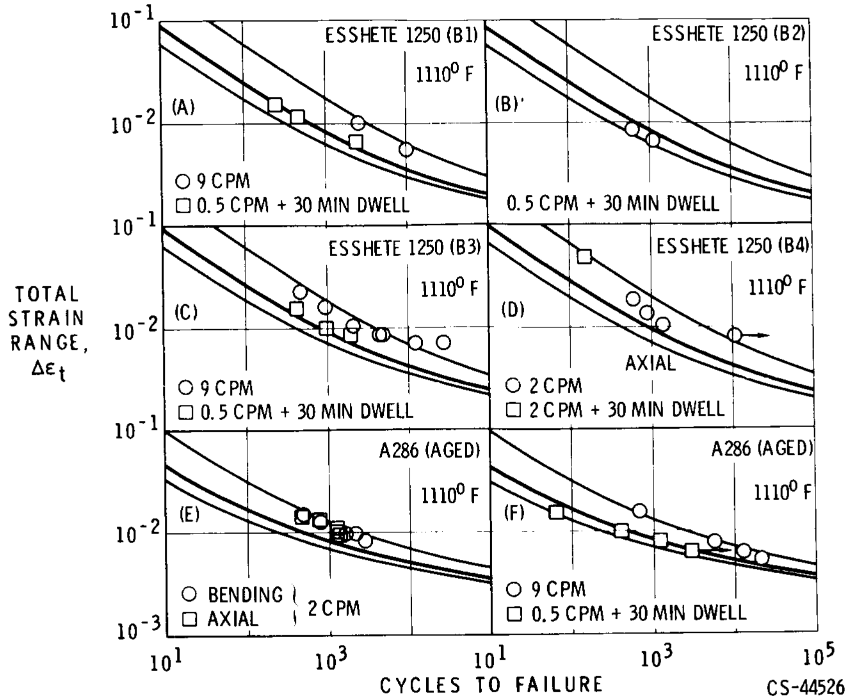

FIGURE 3. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [IRON-NICKEL-CHROMIUM-MOLYBDENUM ALLOYS] (A) - (F) DAWSON, ELDER, HILL, AND PRICE (REF. 4).

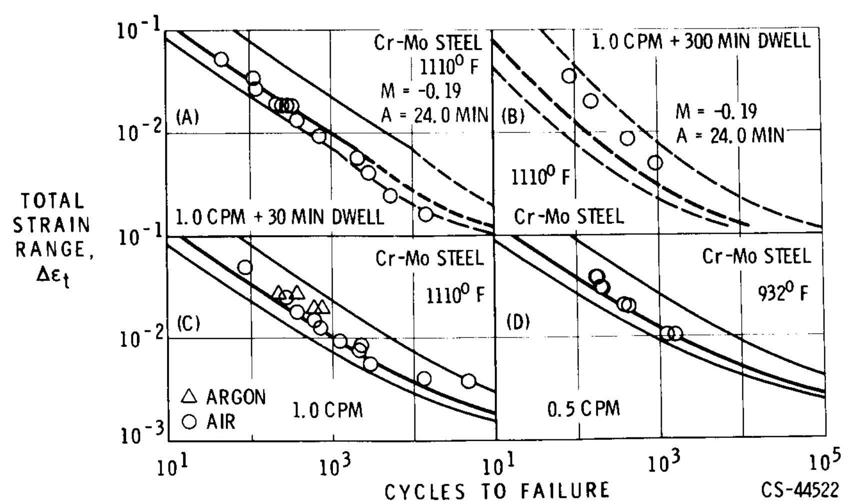

FIGURE 4. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [CHROMIUM-MOLYBDENUM STEELS] (A) - (C) EDMUNDS AND WHITE (REF. 5); (D) JOHANSSON (REF. 6).

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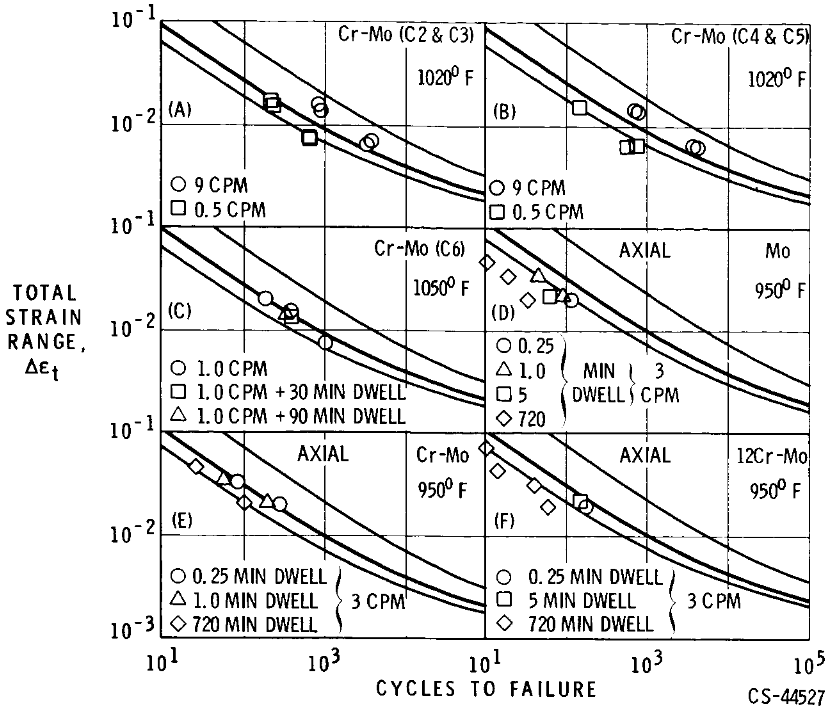

FIGURE 5. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [CHROMIUM - MOLYBDENUM STEELS] (A) - (C) COLES, HILL, DAWSON, AND WATSON (REF. 7); (D) - (F) WALKER (REF. 8).

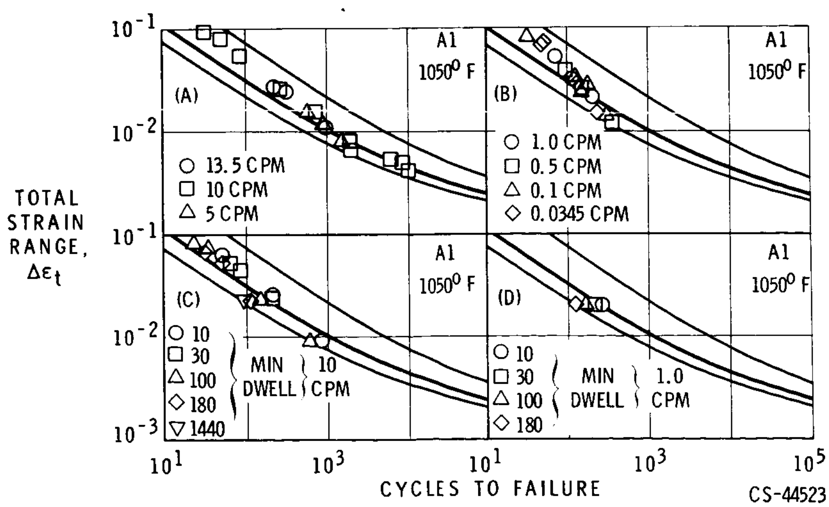

FIGURE 6. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [CHROMIUM-MOLYBDENUM-VANADIUM STEELS] (A) - (D) COLES, HILL, DAWSON, AND WATSON (REF. 7).

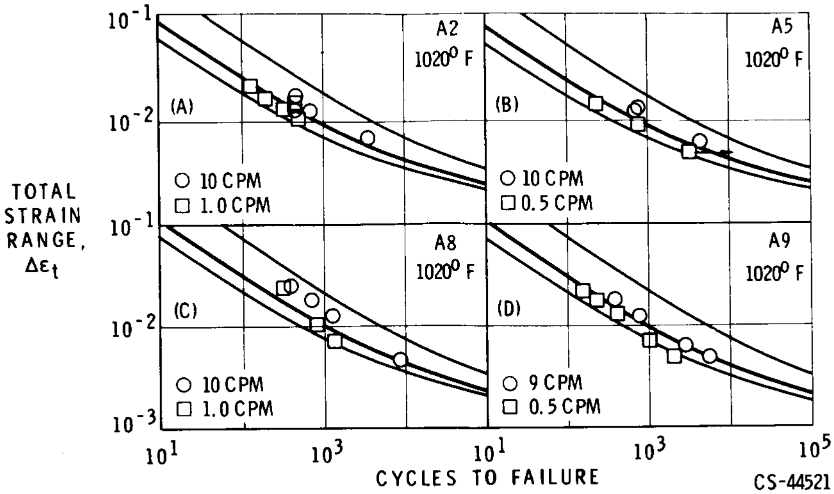

FIGURE 7. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [CHROMIUM-MOLYBDENUM-VANADIUM STEELS] (A) - (D) COLES, HILL, DAWSON, AND WATSON (REF. 7).

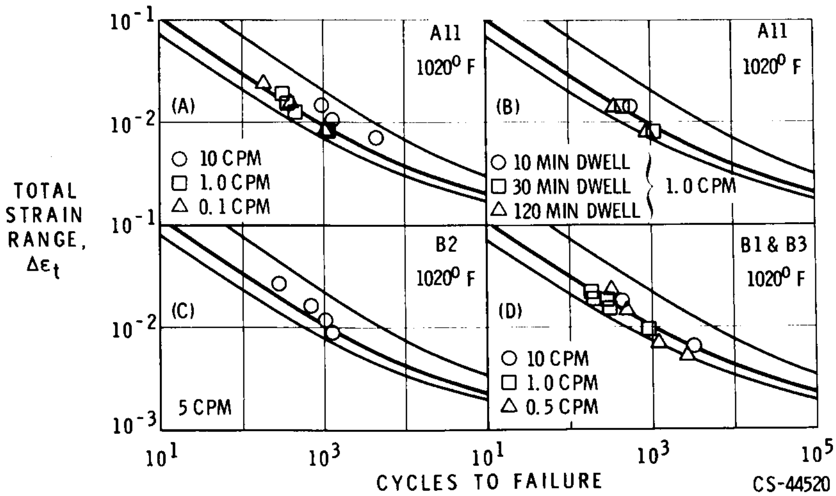

FIGURE 8. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [CHROMIUM-MOLYBDENUM-VANADIUM STEELS] (A) - (D) COLES, HILL, DAWSON, AND WATSON (REF. 7).

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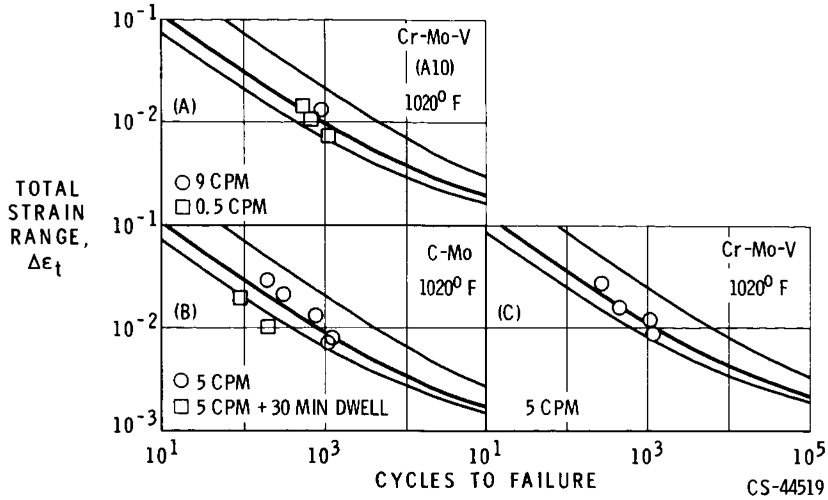

FIGURE 9. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [CHROMIUM-MOLYBDENUM-VANADIUM STEELS] (A) COLES, HILL, DAWSON, AND WATSON (REF. 7); (B) - (C) COLES AND CHITTY (REF. 9).

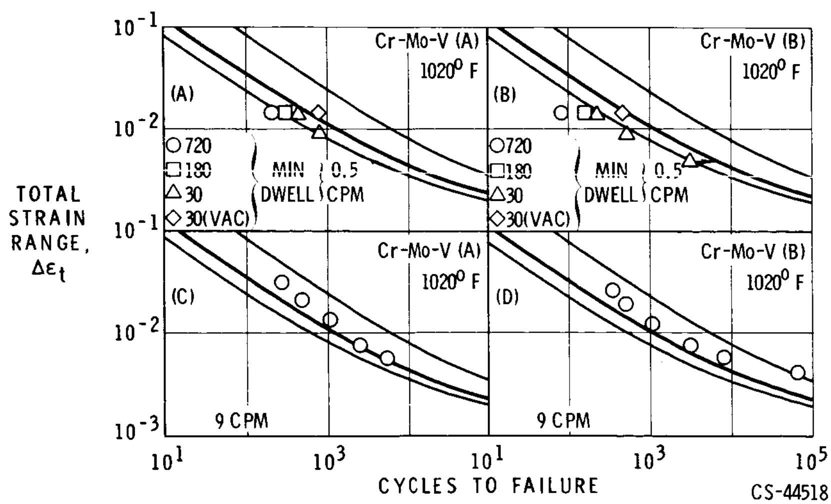

FIGURE 10. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [CHROMIUM-MOLYBDENUM-VANADIUM STEELS] (A) - (D) HILL (REF. 10).

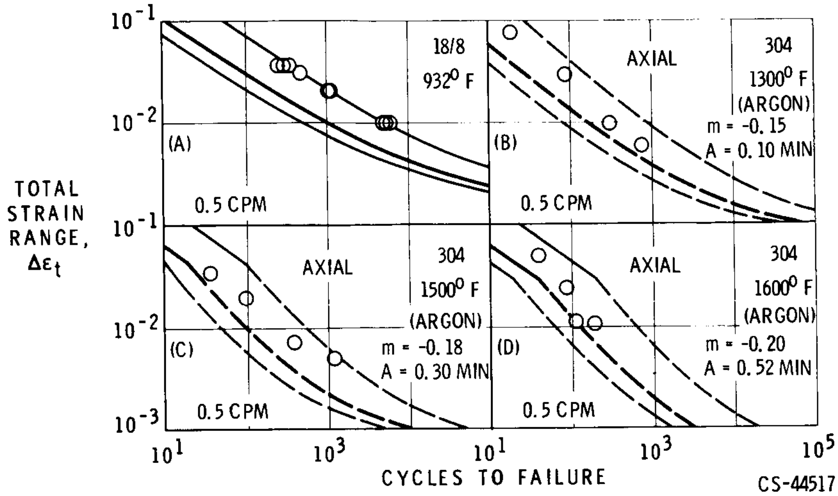

FIGURE 11. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [STAINLESS STEELS] (A) JOHANSSON (REF. 6); (B) - (D) SWINDEMAN (REF. 11).

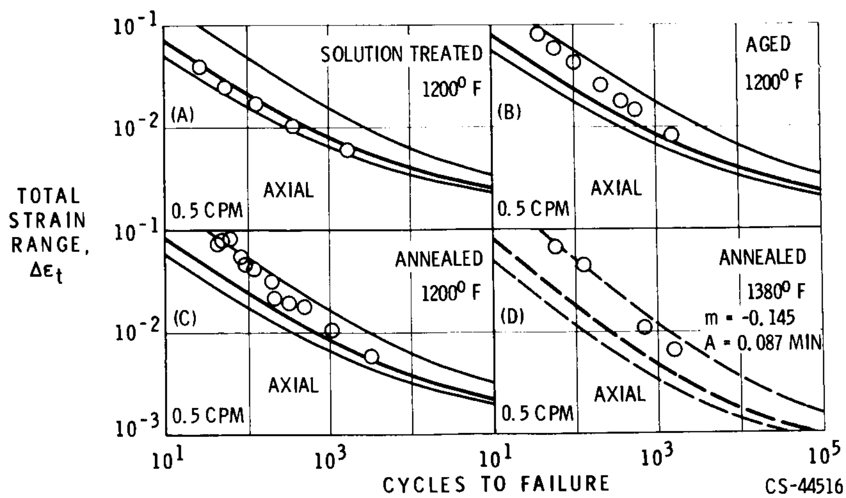

FIGURE 12. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [STAINLESS STEEL (20/25/Nb)] (A) - (D) SUMNER (REF. 12).

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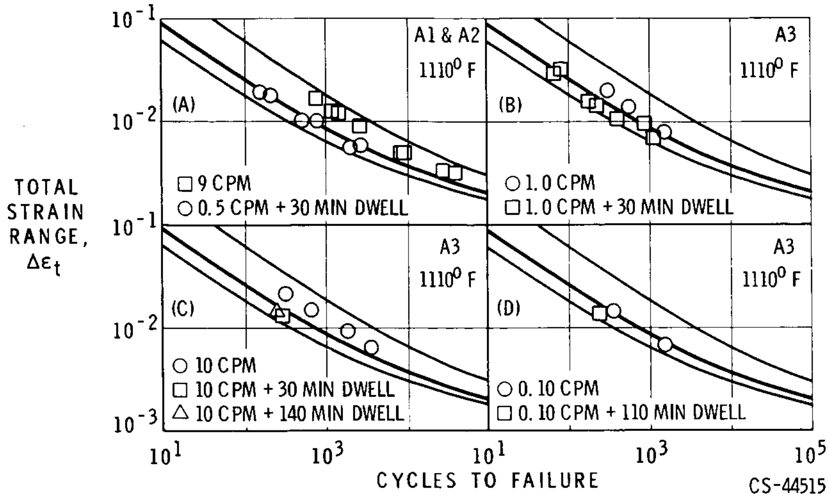

FIGURE 13. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [STAINLESS STEELS (TYPE 316)] (A) - (D) DAWSON, ELDER, HILL, AND PRICE (REF. 4).

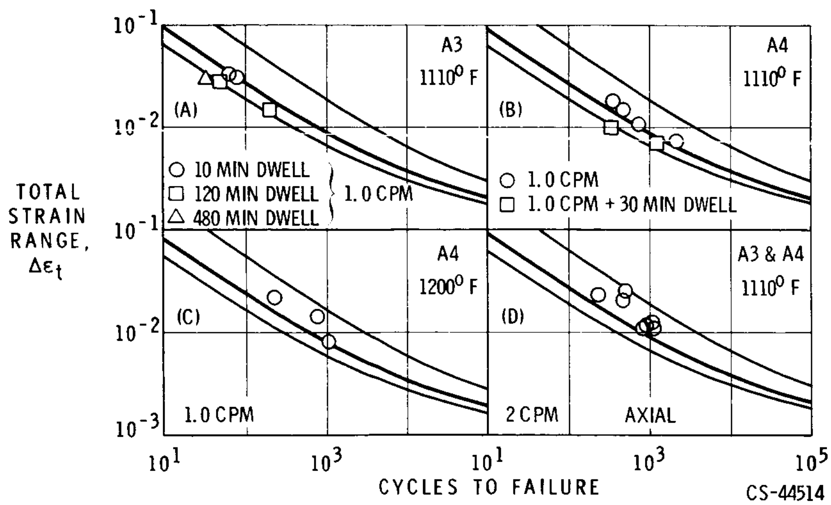

FIGURE 14. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [STAINLESS STEEL (TYPE 316)] (A) - (D) DAWSON, ELDER, HILL, AND PRICE (REF. 4).

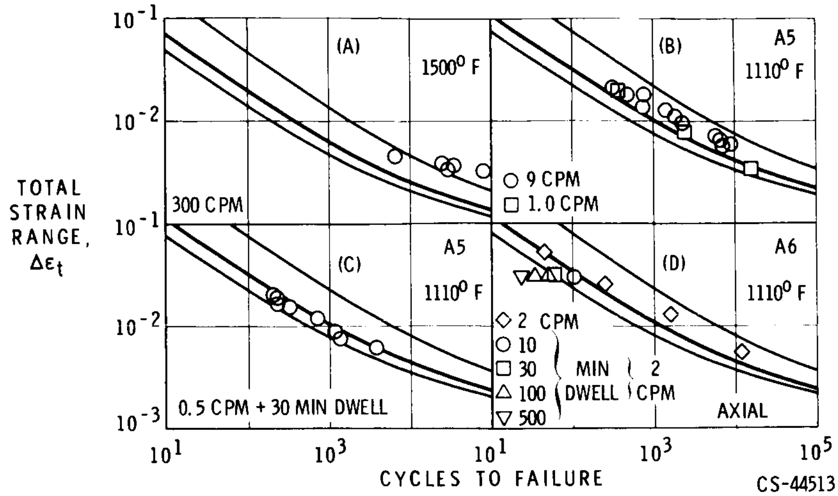

FIGURE 15. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [STAINLESS STEEL (TYPE 316)] (A) DANEK, SMITH, AND ACHTER (REF. 13); (B) - (D) DAWSON, ELDER, HILL, AND PRICE (REF. 4).

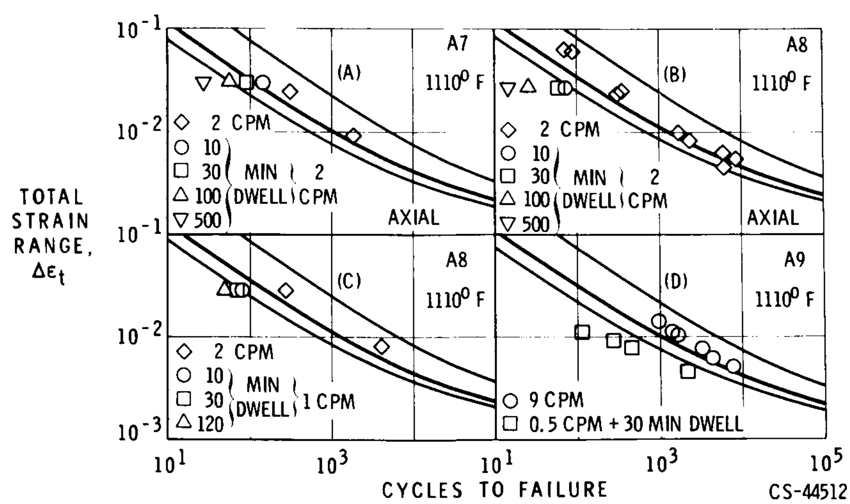

FIGURE 16. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [STAINLESS STEEL (TYPE 316)] (A) - (D) DAWSON, ELDER, HILL, AND PRICE (REF. 4).

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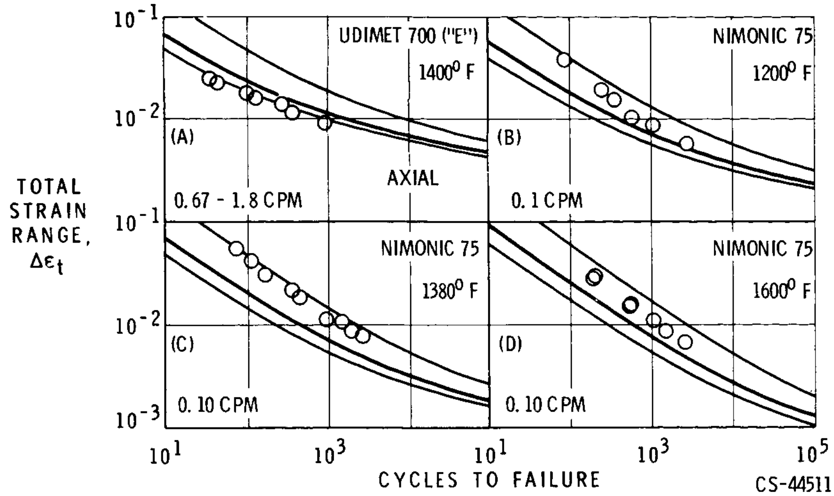

FIGURE 17. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR[NICKEL-BASE ALLOYS](A) WELLS AND SULLIVAN (REF. 14);(B) - (D) FORREST AND ARMSTRONG (REF. 15).

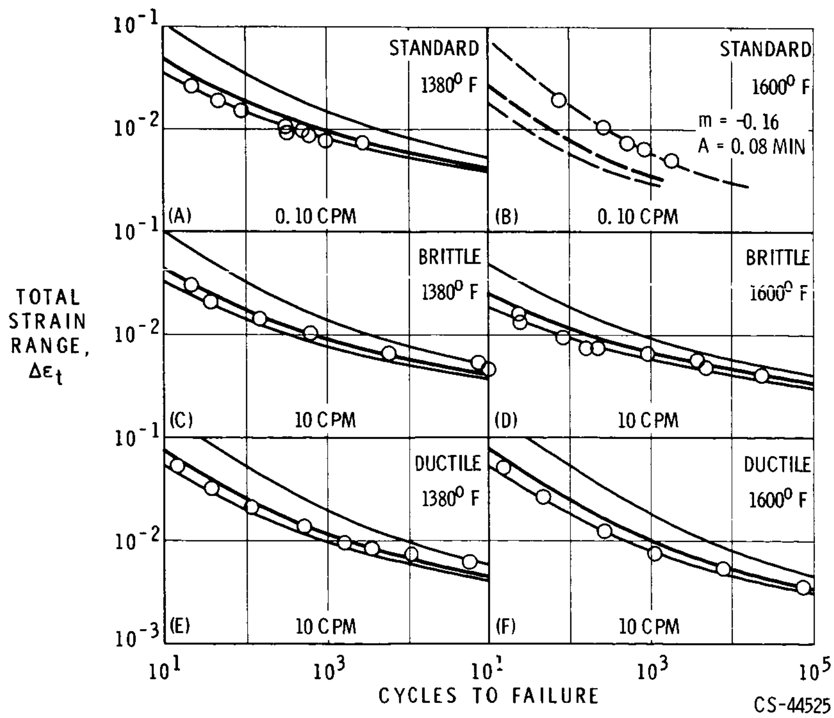

FIGURE 18. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR[NICKEL-BASE ALLOY (NIMONIC 105)](A) - (F) FORREST AND ARMSTRONG (REF. 15).

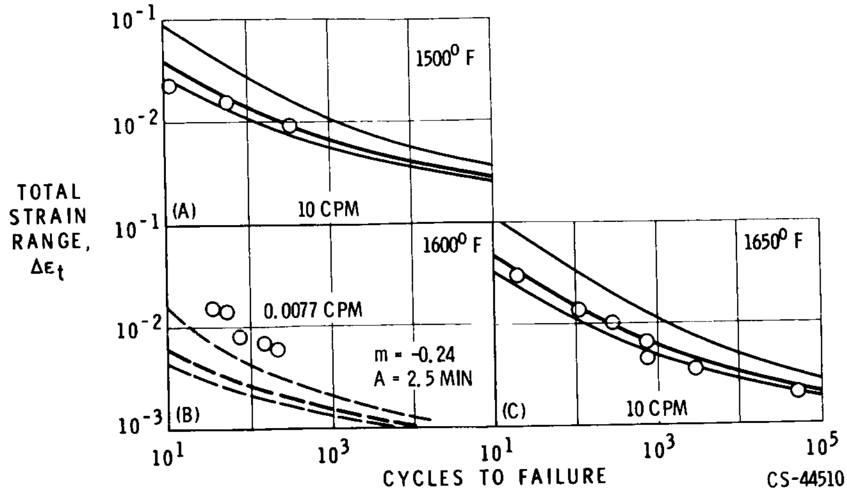

FIGURE 19. - COMPARISON OF ESTIMATED AND OBSERVED BEHAVIOR [NICKEL-BASE ALLOY (NIMONIC 90)] (A) - (C) FORREST AND ARMSTRONG (REF. 15).

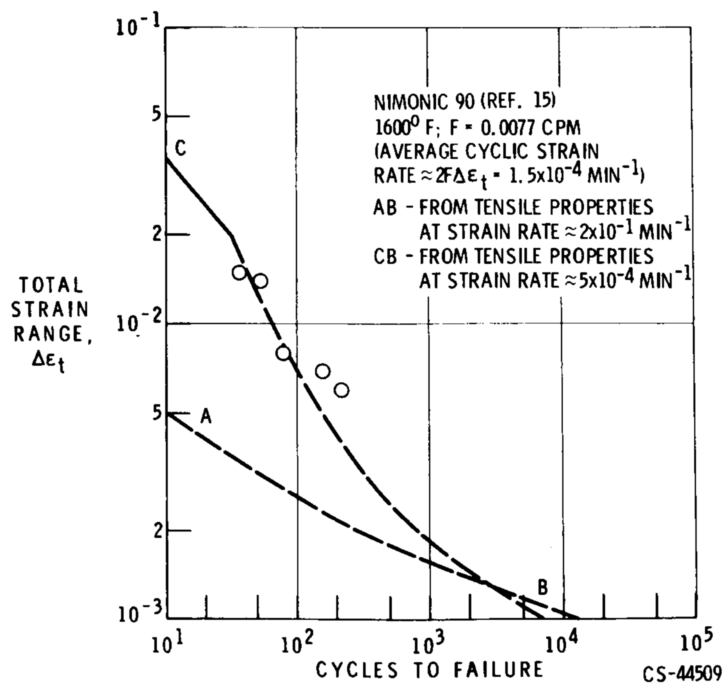

FIGURE 20. - IMPROVED ESTIMATE OF AVERAGE FATIGUE BEHAVIOR BY CONSIDERATION OF STRAIN RATE EFFECTS ON TENSILE PROPERTIES.